A STUDY OF THE EFFECTS OF THE ALPHA TO GAMMA PHASE TRANSFORMATION ON THE SHORT TIME CREEP BEHAVIOR OF AN IRON-CHROMIUM-NICKEL ALLOY

LAWRENCE J. KEILY

A STUDY OF THE EFFECTS OF THE ALPHA TO GAMMA PHASE TRANSFORMATION ON THE SHORT TIME CREEP BEHAVIOR OF AN IRON-CHROMIUM-NICKEL ALLOY

* * * * * * * * * * * * * *

LAWRENCE J. KEILY

A STUDY OF THE EFFECT OF THE ALPHA TO GAMMA PHASE TRANSFORMATION ON THE SHORT TIME CREEP BEHAVIOR OF AN IRON-CHROMIUM-NICKEL ALLOY

by

Lawrence J. Keily

Lieutenant, United States Navy

Submitted in partial fulfillment of
the requirements for the degree of

MASTER OF SCIENCE
IN
MECHANICAL ENGINEERING

United States Naval Postgraduate School
Monterey, California
1958

A STUDY OF THE EFFECT OF THE ALPHA TO GAMMA PHASE TRANSFORMATION ON THE SHORT TIME CREEP BEHAVIOR OF AN IRON-CHROMIUM-NICKEL ALLOY

by

Lawrence J. Keily

This work is accepted as fulfilling
the thesis requirements for the degree of

MASTER OF SCIENCE

IN

MECHANICAL ENGINEERING

from the

United States Naval Postgraduate School

ABSTRACT

An investigation was undertaken in order to determine the effect of the alpha to gamma transformation on the short time creep behavior of an iron-10% chromiun-10% nickel alloy. Test temperatures ranged from 450°C to 600°C, initial stresses ranged from 15,000 to 80,000 psi, and time to rupture ranged from 10 minutes to 180 hours.

Anomalous creep behavior was obtained at 500 and 525°C and was attributed to the alpha to gamma transformation. Rupture elongation was generally found to decrease with increasing stress and decreasing temperatures at all temperatures except at 600°C where the material was in the austenitic condition.

ACKNOWLEDGEMENTS

The writer wishes to express his appreciation for the assistance given by his Faculty Advisor, Alfred Goldberg, Assistant Professor of Metallurgy; C.H.Rothauge, Professor Electrical Engineering; J.A.Clark, Professor of Metallurgy, and other members of the Faculty and Staff of the United States Naval Postgraduate School.

TABLE OF CONTENTS

LIST OF ILLUSTRATIONS

CHAPTER I

INTRODUCTION

In the field of guided missiles a greater knowledge of the short time creep behavior of metals is urgently required. At the high temperatures and stresses which occur during a missile flight, portions of the material of the missile undergo structural changes, which exert a profound influence on the creep rupture characteristics of the missile material. Such changes are called instabilities and include such phenomena as oxidation, recrystallization, grain boundary precipitation, and phase changes. At present, a considerable portion of metallurgical research is endeavoring to discover at what initial conditions a metal should be in order to reduce, if not eliminate, the bad effects, and how to optimize the good effects on high temperature and stress behavior, produced by the occurrence of an instability.

Creep data are generally presented in the form of plots of stress versus rupture time, and stress versus minimum creep rate for various values of the temperature parameter. In the literature the basic creep curve, strain versus time, is seldom presented, and even when available data points are lacking, or if presented, are too few to show the effect of any instability. In the past studies concerning instabilities on creep behavior, the instability of phase change has never been studied without possibilities of other instabilities occurring concurrently.

It was originally planned to use a high purity Fe-25% Ni alloy for the work in this thesis; however, in hot forging, the Fe-Ni ingot suffered severe cracking and was rendered useless. Instead, a Fe-Cr-Ni alloy was used.

In this thesis it was proposed to make an experimental study of the alpha to gamma phase transformation on the creep behavior of a Fe-Cr-Ni alloy.

Part of this study is a careful scrutiny of the basic creep curve to determine the effect of the phase change instability on this curve and on the conventional creep plots mentioned above. Such an approach carried out in an extensive program on numerous alloys at varying stresses and temperatures should be of great assistance in solving the problem of extrapolation of short-time tests to long-time creep behavior.

Temperatures of testing ranged from $400^{\circ} - 600^{\circ}C$, stresses from 25,000-80,000 psi, and time of rupture from 10 minutes to 200 hours. An attempt was made to minimize other instabilities. Oxidation was negligible under $700^{\circ}C$. Recrystallization of the cold worked state was avoided by using an annealed material. The carbon content was sufficiently low to reasonably neglect the effect of carbide growth or precipitation.

CHAPTER II

TEST MATERIAL

The vacuum melted iron nickel chromium alloy was received, in ingot form, from Temescal Metallurgical Corporation. The composition of the alloy, as reported from an analysis conducted by Kaiser Steel at Fontana, California, is given in Table 1.

TABLE I

ANALYSIS OF MATERIAL

Element	Percent	Element	Percent
C	.03	Cu	.002
Mn	.01	V	.003
P	.003	Mo	.003
S	.010	Al	.002
Si	.01	Co	.045
Ni	10.6	Fe	remainder
Cr	10.6		

For use in the experiments performed in this thesis, an homogeneous structure was required. If the material was used in the unhomogenized state, differences in chemical composition between specimens would, in effect, make each specimen a different alloy, with different properties. In order to produce this homogeneous structure, experience indicated that this particular alloy must be heated for a period of three to four weeks at temperatures near to its melting point (approximately 1420°C). The ingot was slowly heated from 1,000 to 1400°C over a period of eleven days and remained at 1400°C for seven days. The treatment was stopped at this time, because the heating elements of the furnace burned out. To prevent oxidation while heating, a vacuum furnace was used, the vacuum being maintained at 10^{-5} mm Hg by means of a diffusion pump.

Upon completion of the homogenization process the material was hot forged to approximately ½-inch thickness. It was then reduced by rolling to a strip .09" X 1-3/4" X 7ft. The rolling process was accomplished in two steps. First, the material was rolled to ¼-inch thickness and it was ascertained that the hardness of the material had increased to such an extent as to require annealing before the material could safely be further reduced. To be softened the material was annealed for 24 hours at 750°C and permitted to furnace cool. Next, the surface was cleaned of a slight oxide film formed during heating, by pickling with dilute hydrochloric acid. Finally , the material was rolled to its final dimensions.

Thirty-six specimens were machined from 4¼-inch blanks cut from the above strip. The nominal dimensions of the specimens are shown in Fig.1, and variations of ± 0.001" were obtained in the reduced section of the finished specimen.

After machining the specimens were slowly heated to 750°C, annealed for one hour at temperature, and then furnace cooled. The material was then cooled in liquid nitrogen (melting point, -195°C) for twelve hours, to cause any austenite present to transform to the ferrite phase. Subsequent metallographic and x-ray examinations of the cooled material did not reveal the presence of the gamma phase.

One of the more time consuming problems encountered was the determination of the temperature of the alpha to gamma transformation. An examination of the equilibrium diagrams for Fe-Cr-Ni alloys established that, for this composition, alpha phase begins to transform on heating at approximately 375°C and continues to do so until 600°C.(1) However, the Fe-Cr-Ni system is well known for its sluggishness in reaching equilibrium (2), so that, at the lower temperatures used in this experiment, the material

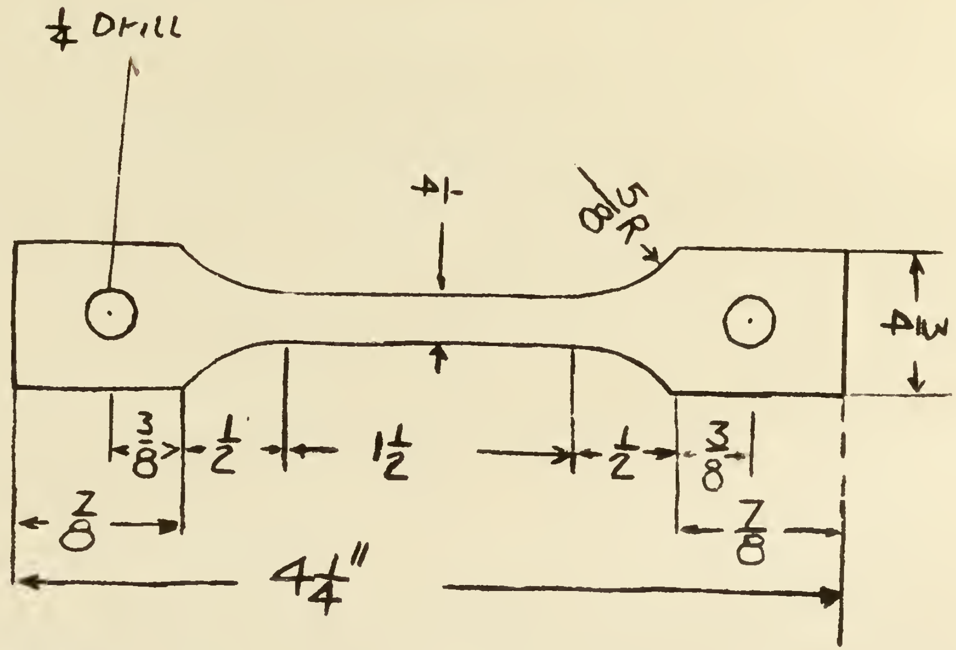


FIG. 1 DIAGRAM OF SPECIMEN

would seem far from equilibrium; therefore, on the basis of this information and data presented subsequently, the material is presumed to be predominantly of the alpha phase below 550°C and of the gamma phase above 550°C.

From a dilatometric study of the alloy, conducted by heating a creep specimen from room temperature to about 700°C at rates varying from $\frac{1}{2}$ to 2°C per minute, the following data on the transformation was obtained: On run #1, the material transformed between 600-610°C. On run #2, approximately 30% of the material transformed at 490°C and the remainder at 600°C. On run #3 and #4 the material transformed at 550-560°C. The change in length in a one-inch gage length when the specimen transformed was approximately a contraction of 0.005". In addition, a creep specimen was heated for 2.5 days at 500°C and no transformation was observed. A plot of results, obtained from one of the dilatometric runs, is illustrated in Fig.2.

An attempt was made to ascertain when transformation occurred by heating the material and measuring the resistance change with temperature by means of a Kelvin bridge. By an examination of a cartesian plot of resistance versus temperature, it was hoped to be able to tell when a phase change took place. In general, as the temperature of a metal increases the resistance also increases; also, it is known that the resistance of austenite is greater than that of ferrite; hence, an increase in $\frac{dR}{dT}$ should occur when transformation takes place; and upon completion of transformation $\frac{dR}{dT}$ should decrease in value . The curves obtained did not exhibit a pronounced change in slope, so that, the temperature of the phase change could not be detected by this method. The above method has been successfully used several times in the past in determining the equilibrium temperature of alloys.(3) All of the alloys tested were of different composition from the alloy used in this work.

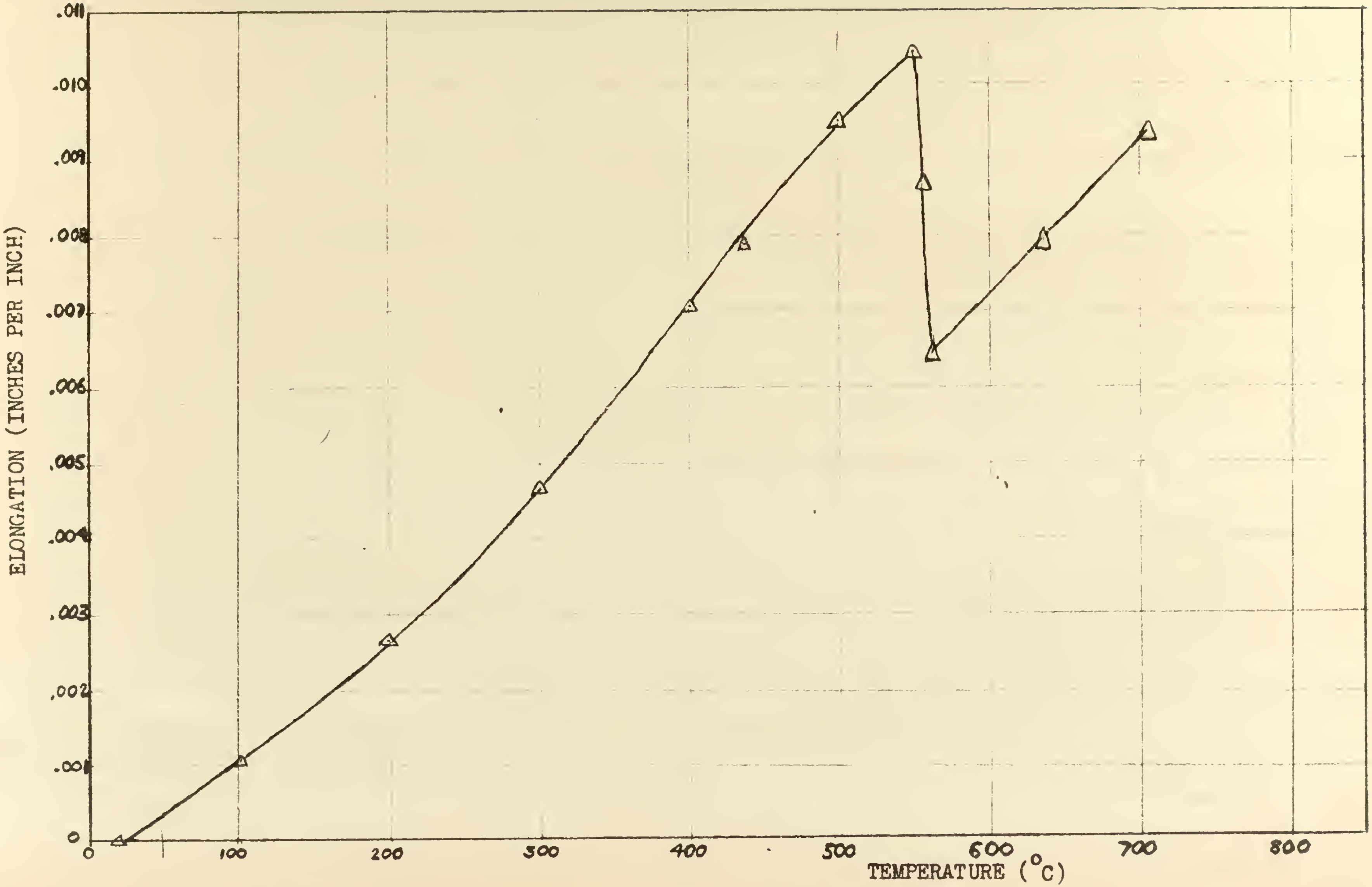


FIG. 2 DILATOMETRIC RUN CONDUCTED AT A RATE OF HEATING OF 1°C PER MINUTE

CHAPTER III

EXPERIMENTAL EQUIPMENT

A conventional constant load-single lever type testing machine was used. The lever arm rested on two case-hardened knife edges. The load was applied to the weight pan on the rear end of the lever arm and transmitted through the pulling tab and extensometer rod to the specimen.

The upper swivel block was connected to the lever by a 5/8" pin located at right angles to the knife edges. Attached to the swivel block was a turnbuckle, used for adjusting the height of the lever arm. Connected to the turnbuckle was a double universal joint. This joint along with the lower universal joint ensures that the system has three degrees of freedom. Following the upper universal joint came the following components interconnected to one another in the order listed: upper pulling tab, the specimen, lower pulling tab, lower universal joint, and lower turnbuckle which was secured to the base of the testing machine.

For purposes of balancing the lever system, before applying a creep load, a moveable lead counterweight was located on a screw extending from the front of the arm. The axis of the screw was perpendicular to the axis of the knife edges.

The exact lever arm ratios (calibrated with lever arm horizontal and also in extreme upper and lower positions), had been previously determined by placing a Baldwin SR-4 type (0-2000 lb.) load cell in series with the pulling tab. Checking the readings against calibrated weights, a 14.71 to 1 ratio for creep unit #5, and a 9.98 to 1 ratio for unit #4 were the values determined.

Conventional tube-type nichrome wound furnaces were used on the creep unit. An L and N temperature recording controller in conjunction with a

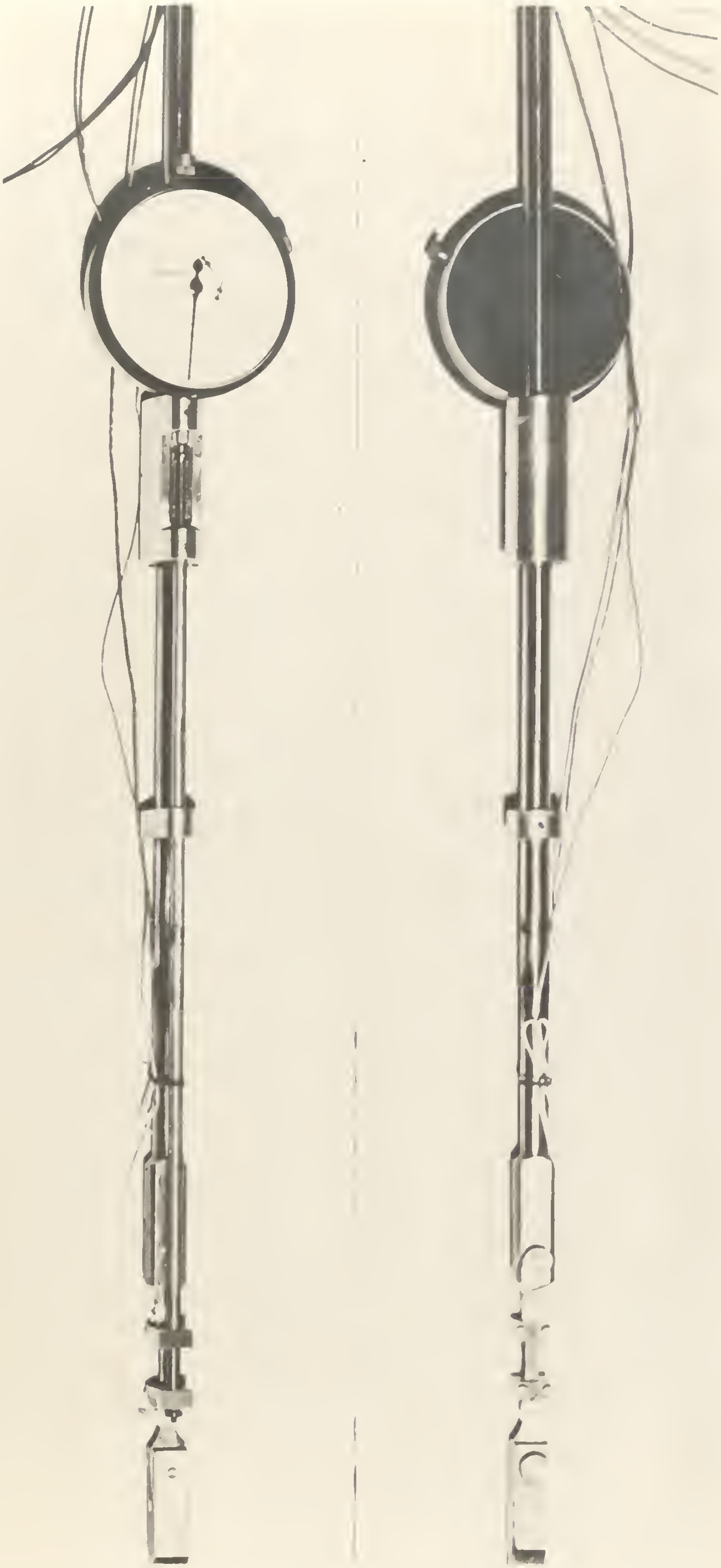
Duration Adjusting Temperature control unit held the temperature to within $\pm 2^{\circ}C$.

CHAPTER IV

EXPERIMENTAL TECHNIQUE

First, the thickness and width of the specimen were measured at three successive locations, in the reduced gage section of the specimen, to the nearest 0.0001" , by means of a micrometer. The area was computed at each of these three locations and the smallest ares was used to determine the load required to give the desired initial stress. The specimen was attached to the upper pulling tab by means of a $\frac{1}{4}$-inch pin. Next, the gage blocks, which had 1/8-inch stellite indentation points, were located 1" $\pm$0.001" apart in the reduced gage section of the specimen by means of a jig specially made for this purpose. Then, the lower pulling tab and a thermocouple, tied to the narrow region, were attached. A Starrett gage having a minimum division of 10^{-4} inches with a total extension of 0.4 inches was used to determine gage section elongation. A picture of the specimen, extensometer gage, rods and blocks, is shown in Fig.3.

The above rig was placed into the air furnace, which was at test temperature, and the upper and lower pulling tabs were pinned to their respective universal joint connections. Approximately 20-30 minutes were required for the specimen to reach temperature equilibrium. After equilibrium was established, an hydraulic jack was placed under the weight pan and the load placed on the pan. Next, the specimen was gradually loaded by the lowering of the jack. Time of loading was less than 30 seconds for all specimens. Periodic temperature and extensometer dial readings were taken. The time of rupture was automatically recorded, the falling of the creep machine lever opening a microswitch when the specimen broke.



Extensometer Unit Figure 3

CHAPTER V

THEORETICAL ASPECTS

PART I: A DISCUSSION OF THE BASIC CREEP CURVE

The conventional basic creep curve is composed of three successive regions, namely the primary, secondary, and tertiary. In the primary region, deformation increases with a decreasing rate. Following the primary is the secondary region in which a minimum creep rate is obtained, and finally the tertiary region in which the creep rate accelerates until rupture. The fracture may be either transcrystalline or intercrystalline. A transcrystalline fracture is essentially ductile, the slip mechanism being slippage of planes of atoms within a crystal with respect to one another. An intergranular fracture is brittle and an examination of the microstructure reveals voids and cracks along the grain boundaries. A transcrystalline failure is more likely to occur at low stresses and temperatures; whereas, an intergranular failure is more likely to occur at high stresses and temperatures. In all instances reviewed in the literature, at elevated temperatures, elongation at rupture decreases as load decreases.

The irregularities in the basic creep curve appear most clearly as an accelerated rate in the primary or the secondary stages. This is referred to as a "false tertiary" because, the acceleration is followed by a new steady state creep rate and finally followed by the true tertiary, leading to rupture. The above phenomenon was observed in several papers (4,5) in which its appearance was attributed to recrystallization, occurring during creep. The phenomenon observed in creep testing of aluminum was attributed to precipitation.(6) However, it is to be emphasized, that a false tertiary is not necessarily the only way in which an instability may manifest itself. For instance, Smith, who observed this tertiary in specimens under low

stress, attributed its occurrence to recrystallization; however, at higher stresses where recrystallization was also observed to occur, a false tertiary was not observed. (7)

In metals, atoms strive towards the lowest state of free energy. At a given temperature the gamma phase is at a lower free energy state than the alpha phase. However, before the atoms of a metal can move from their positions in the alpha matrix to those in the gamma matrix, absorption of a certain amount of activation energy is necessary. At a certain minimum temperature the metal will transform spontaneously by the application of heat alone, call this temperature T_o. Below T_o activation energy can not be supplied by heat alone. This deficiency of activation energy may be compensated for by stress and the material made to transform at temperatures below T_o. A number of studies have been made showing that this is actually the case, such as the study made by Kaufman and Cohen.(3) Fig.4,copied from the work by Kaufman and Cohen,illustrates the effect of stress on phase changes in the Fe-Ni system.

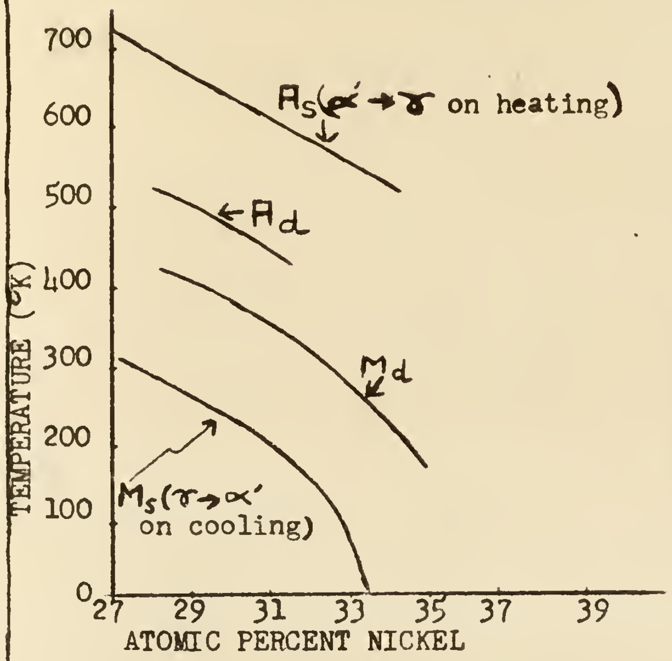
Of further bearing on any theoretical study of the phase change instability on the basic creep curve are the facts, that most phase changes are essentially diffusion processes,and that as temperature increases, the rate of diffusion does likewise.

PART II: PLOTTING METHODS:

In attempts to develop mathematical formulae for the creep variables, various schemes have been used. Among these are the following equations, the first by Machlin and Nowick(8):

$$\log t_r = \frac{A + BT - Ds}{T}$$

where t_r = time to rupture
s = stress
T = absolute temperature
A,B,and D = constants



EXPLANATION OF SYMBOLS:

A_s- temperature at which transformation from $\alpha' \to \gamma$ occurs on heating

A_d- lowest temperature at which the $\alpha' \to \gamma$ reaction can be initiated by deformation

M_s- temperature at which transformation from $\gamma \to \alpha'$ occurs on heating

M_d- lowest temperature at which the $\gamma \to \alpha'$ reaction can be initiated by deformation

KAUFMAN AND COHEN (3)

Fig. 4. The effect of stress on $\alpha' \to \gamma$ and $\gamma \to \alpha'$ phase changes over a range of composition for the Fe-Ni system.

The other equation derived by Kauzmann (7) and Duchman (8):

$$\log \dot{e}_{min} = \log r + Ps$$

where r and P are constants, at constant temperature
$\dot{e}_{min}$ = minimum creep rate
s = stress

The most important observation of these two equations is that a straight line should be obtained when stress is plotted against the logarithm of the minimum creep rate or against the logarithm of time to rupture for a given test temperature.

It has been shown in several papers, for instance (11,12,13), that semilogarithmic plots of the above type do not yield straight lines even in the absence of instabilities. Concerning this method, Grant (14) states that "Aside from the fact that there is disagreement between theory and experimental results, the method does not appear to offer much hope as a means of interpolation or extropolation of creep rupture data. The opportunities for decreasing the amount of testing or of shortening the longer time tests do not look attractive."

The method of handling creep rupture data in the present work is essentially that of Grant and Bucklin (15). Stress versus rupture life and stress versus minimum creep rate are plotted on log-log paper. The authors suggested that each material instability caused a deviation from a straight line resulting in an increased slope of the curve. If straight line segments are fitted to the points, sharp breaks in the log-log plots are the result. Neither Grant and Bucklin nor the author of this paper suggest that a sharp transition or break did, in fact, occur at a specific stress or strain rate at a constant temperature, but utilized the sharp breaks as a plotting convenience to optimize the temperature-stress-rupture life or

creep rate relationship for the instability. Each material instability which occurs for one combination of temperature and time is expected to occur at a shorter time at a higher temperature.

Again quoting Grant (14), he states that, "the principal advantage of this graphical method of using creep-rupture data for interpolation and extrapolation is the long saving in long time tests. Once an instability is encountered at some lower temperature, it can be found at shorter times at higher temperatures. Thus, by running more short-time tests (under about 200 or 300 hours), but over a wider range of temperatures, much can be determined regarding the stability or the nature of the instabilities of the material."

CHAPTER VI

EXPERIMENTAL RESULTS AND DISCUSSION

Creep data were obtained at 450, 475, 500, 525, 550, and 600°C for stresses ranging from 25,000 - 80,000 psi.

Log-log plots of initial stress versus rupture time and initial stress versus minimum creep rate are given in Fig.5. A plot of stress versus temperature for minimum creep rates of 0.1% and 1.0% per hour is shown in Fig.6. In Fig.7 is presented a plot of elongation at rupture versus stress. The basic creep curve and comments on these curves are included in Appendix I. Appendix II shows the various initial stresses for each test temperature, the time to rupture, minimum creep rate, total elongation at rupture, and total primary and secondary time.*

Several salient features of the above plots are believed to be associated with the alpha to gamma phase change. In the following, these features are first pointed out and, then later, interpreted.

Referring to Fig.5, it is observed that the log-log plot of initial stress versus rupture time consists of a single straight line segment for temperatures 450, 475, and 600°C; whereas, the curves for 500°C, and 525°C are each composed of two straight line segments intersecting at times 1.3 and 2.7 hours respectively. Considerable scattering is shown at 500°C between 35,000 and 45,000 psi. It further appears that two straight line segments might better represent the data at 475°C, but a single straight line is drawn on the basis of the plots discussed subsequently.

In Fig.5 it is shown that the log initial stress versus log minimum creep rate, plots at 450, 475, and 600°C are composed of a single straight

* Note: On the plots in this thesis, a dotted line is used to indicate uncertainty on how a particular curve progressed.

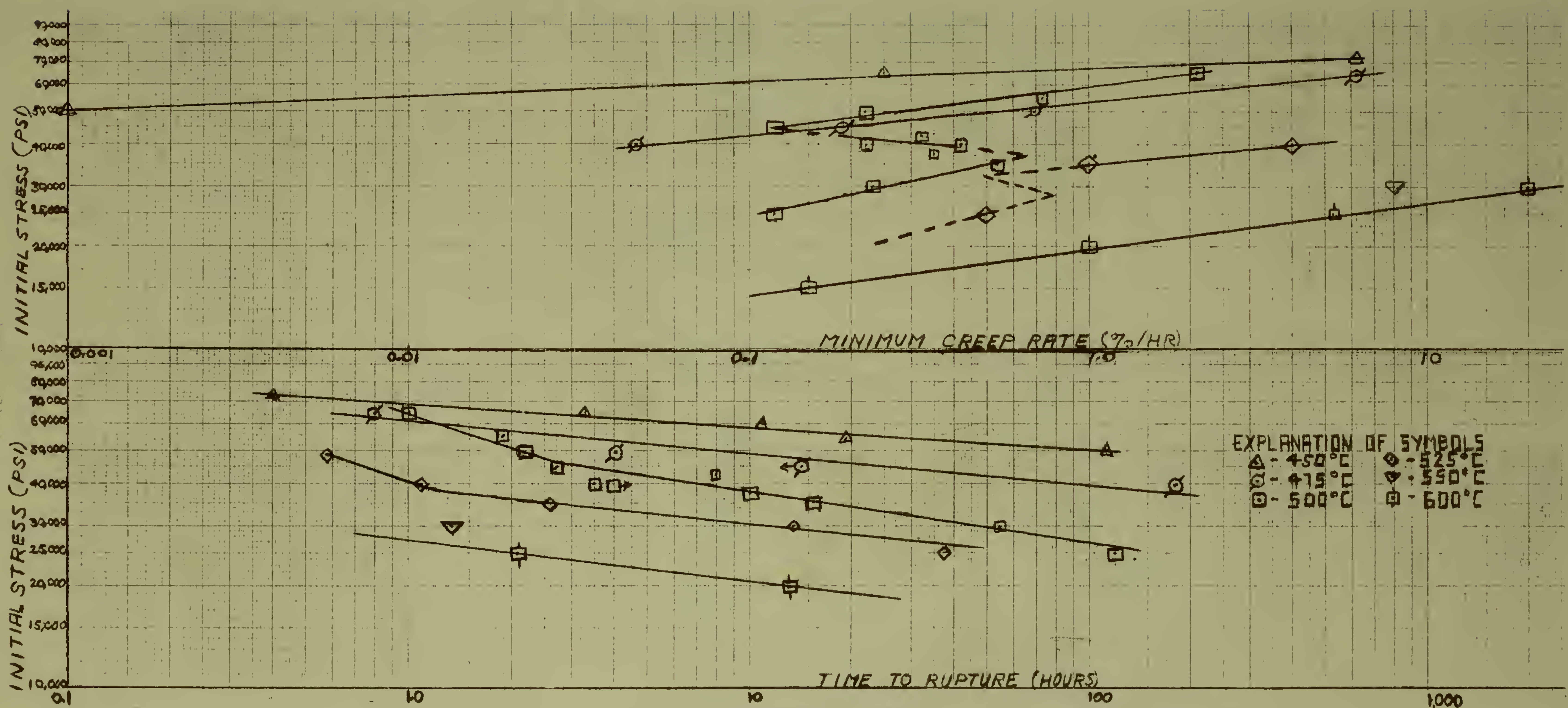


FIG 5 PLOT OF INITIAL STRESS VERSUS RUPTURE TIME AND MINIMUM CREEP RATE

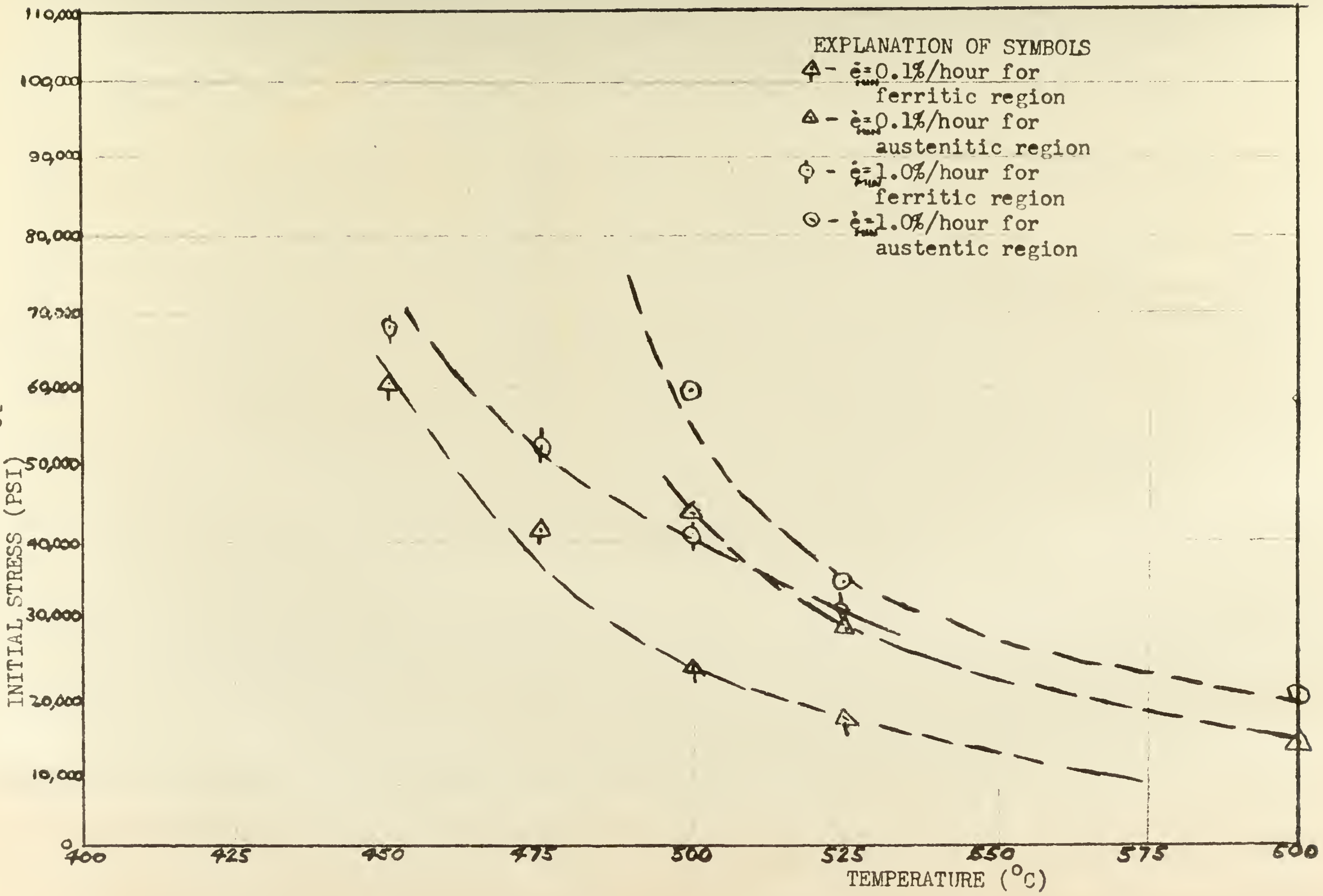


FIG. 6 PLOT OF INITIAL STRESS VS TEMPERATURE FOR MINIMUM CREEP RATES OF 0.1 AND 1.0 PERCENT PER HOUR

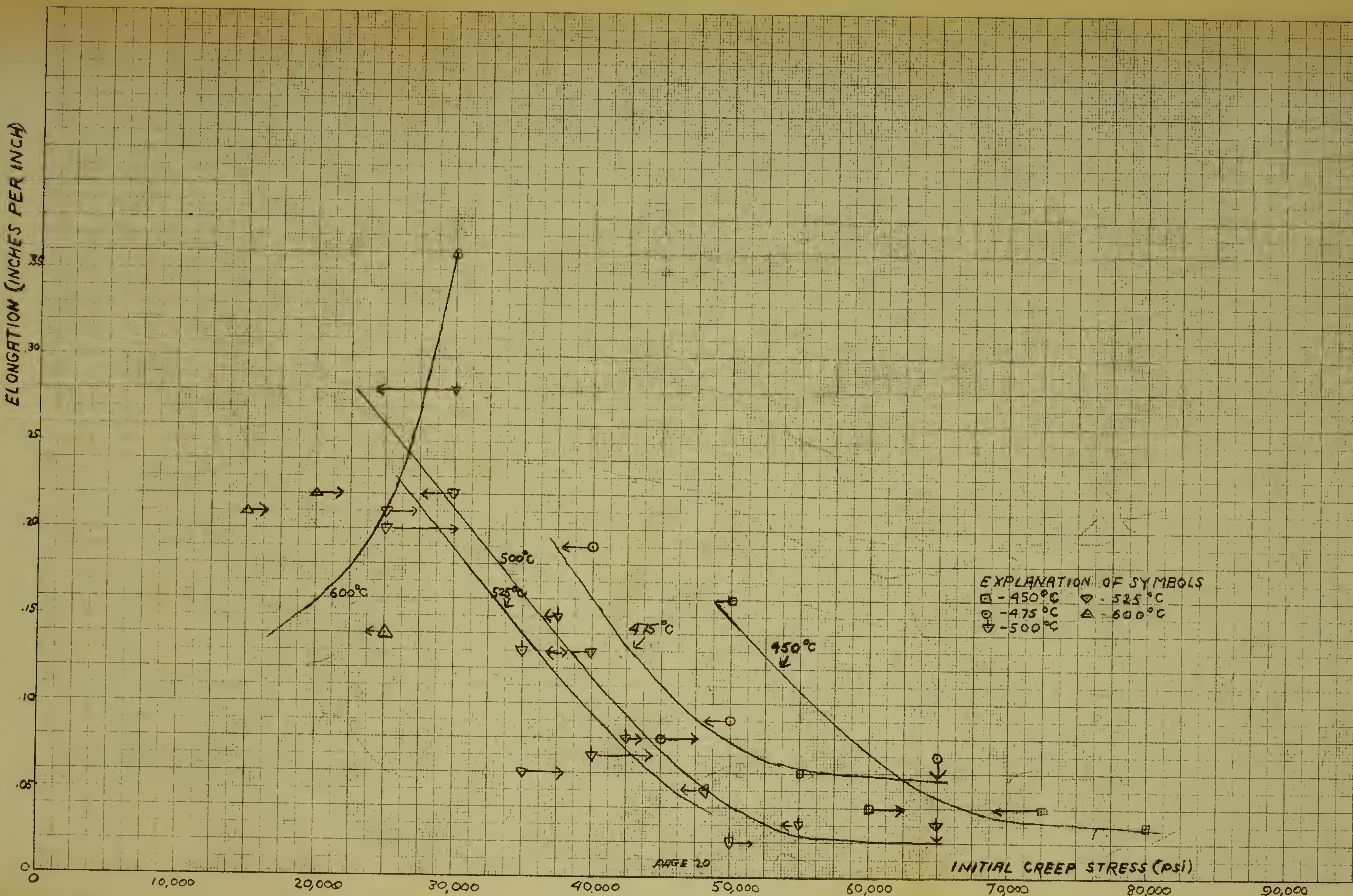


FIG. 7 PLOT OF ELONGATION AT RUPTURE VERSUS INITIAL CREEP STRESS

line segments, having a positive slope; whereas, irregularities exist at 500°C and 525°C. At 500°C the plot appears to be definitely composed of two positive slope straight line segments widely separated from each other and joined by a negatively sloping intermediate line. At 525°C the above occurrence appears to be less definite, as only three points have been obtained with which to construct the curve. However, these three points do not lie in a straight line. Since a rather large angle is formed at the intersection of two straight lines drawn respectively through the 25,000 and 35,000 psi points, and through the 35,000 and 45,000 points, it would seem more logical to draw the 525°C curve similar to that at 500°C.

The basic creep curves for 450, 475, and 600°C are all observed to be conventional. At 500°C, conventional curves are obtained at all stresses up to 39,000 psi, and at 55,000 and 65,000 psi. Curves with extremely short primaries were observed at 40,000 and 42,500 psi. At 45,000 and at 50,000 psi no primary stage was apparent. At 525°C a conventional curve is observed at 25,000 psi and only a tertiary at 35,000 and 40,000 psi. False tertiaries were not observed in any creep runs performed.

The above features are interpreted as follows: The specimens remain ferritic for stresses tested at 450°C and 475°C. Supporting this view at 450°C are the facts that the plots of log stress versus log $\dot{\epsilon}_{min}$ and log rupture life are single straight lines and that only conventional basic creep curves are observed. Concerning the log stress versus log time plot for 475°C, although a single line was drawn, two segments intersecting at 10 hours may have equally well represented the indicated points. The single straight line was chosen because of the minimum creep rate plot being similar to the 450°C and as a result of the information presented below in the load stress versus temperature plot.

At 500°C, at stresses below approximately 38,000 psi the material is ferritic. For these stresses the basic creep curves are conventional and the log stress versus log $\dot{\epsilon}_{min}$ has a positive slope. Between 38,000 and 50,000 psi the material is believed to commence transformation to austenite. At the lower stresses the transformation is probably incomplete. As stress increases the amount transformed increases until finally 100% is transformed. In this stress range the basic creep curves exhibit extremely short primaries at 40,000 and 42,000 psi, and at 45,000 and 50,000 psi no primary. A break in the log stress versus log time curve occurs at 46,000 psi and 2.7 hours; such a break, in the literature, is associated with an instability. Likewise, the break in the log $\dot{\epsilon}_{min}$ versus log stress indicates an instability. Similar scattering in values of elongation was observed by M.Herman and N.Brown (16) in a paper on the influence of an order-disorder reaction on creep behavior at temperature and stresses where the order-disorder reaction is known to occur. At stresses above 50,000 psi the material is believed to transform almost instanteously with the application of the load. In this stress region conventional creep curves were observed. The upper straight line segment on the log $\dot{\epsilon}_{min}$ versus log time plot is believed to be associated with the austenitic region in contrast with the lower which is associated with the ferritic region, although data is scarce.

Similar behavior is believed to occur at 525°C. The lower dotted straight line is associated with the ferritic region, the upper solid line, with the austenitic region. At 35,000 and 40,000 psi, which stress values are in the austenitic region, the basic creep curves are continually accelerating. This behavior is similar to that at 45,000 and 50,000 psi for 500°C.

At stresses at 600°C the material is believed to transform almost instantaneously as only conventional basic creep curves are obtained. The log

stress versus $\dot{\epsilon}_{min}$ and log time plots are linear.

Several ruptured specimens were examined metallographically with the hope of determining any differences in the structure of specimens that had been subjected to different test conditions. Results of this examination revealed no marked difference in appearance. This was to be expected as any transformed austenite in this alloy largely transforms back to ferrite before reaching room temperature.

The plot of initial stress versus temperature further illustrates the above interpretation. This plot was obtained by reading the value of stress directly from the log stress versus log $\dot{\epsilon}_{min}$ for two arbitrary values of $\dot{\epsilon}_{min}$ of 0.1%/hour and 1.0%/hour. The value of stress for 500°C and 525°C was obtained by first extrapolating the austenitic and ferritic regions of these curves and then reading the stress value for both the austenitic and ferritic regions. It can be observed that the austenitic type structure is stronger than the ferritic, as a greater stress is required to produce the same minimum creep rate in the austenitic region.

The rupture strains as a function of initial stress at the different temperatures are shown plotted in Fig.7. The results show a general decrease in strain with increasing load and decreasing test temperature,both in contradiction to what is generally obtained. However, the 600°C data does show a general increased rupture strain with stress. The reversed trend obtained at the lower temperatures may well be the result of the alpha to gamma change. If this be true, the conventional behavior should then be obtained below some lower temperatures where transformation is not induced by stress. A material which has a short rupture elongation at elevated stresses and temperatures may be useful in the missile field.

CHAPTER VII

CONCLUSIONS AND RECOMMENDATIONS

On the basis of experimental data obtained, the following conclusions are drawn for the Fe-Cr-Ni alloy used: 1. It appears that in the temperature region where the phase change does not occur, regardless of how large a stress is applied, that the basic creep curve is conventional. Further, the log stress versus log rupture time and log stress versus log minimum creep rate appear linear.

2. In the temperature region where the phase change may be induced by stress the following effects appear to occur: At stresses too low to induce a phase change, the basic creep curve is conventional. At the lower values of stresses that induce the phase change the basic creep curve exhibits a very short primary or none at all. At the higher transformation stresses, the basic creep curve is again conventional. No false tertiaries appear to occur. A discontinuity is observed in the log stress versus log rupture time plot. The log stress versus log minimum creep rate appears to be composed of two positive-sloping widely separated straight line segments, the lower segment representing the ferrite region and the upper segment the austenite. These two segments are joined by a linear negatively-sloped segment.

3. In the temperature region above T_0 where the phase change may occur by heat alone, the basic creep curve is conventional. Log stress versus log rupture time and log stress versus log minimum creep rate are conventional.

4. Elongation at rupture decreased with increasing stress at temperatures below the T_0 transformation temperature. At temperatures above this temperature elongation increases with stress.

The following recommendations are proposed concerning performances of future theses on phase change instabilities: 1. Conduct extensive dilatometric surveys of the phase change characteristics to obtain knowledge of the percent transformed as a function of time at various temperatures.

2. A survey of the minimum stress at a given temperature and the lowest temperature at which transformation is induced by stress.

3. A detailed study of the relations between fracture, strain, temperature, and stress covering the alpha, alpha and gamma, and gamma regions.

4. It is recommended that instrumentation be used to automatically record elongation and time. At present, only one such instrument is available at the United States Naval Postgraduate School.

BIBLIOGRAPHY

1. J.H.Monypenny, Stainless Iron and Steel, Vol.II, Chapman and Hall, Ltd, p.157,1954.

2. R.H.Aborn and E.C.Bain,"Nature of Nickel-Chromium Rustless Steels," Trans. ASST, Vol.18, p.837, 1930.

3. Larry Kaufman and Morris Cohen, "The Martensitic Transformation in the Iron-Nickel System", Journal of Metals, Transactions Section,Oct. 1956,p.1393.

4. B.Felton, Proc.Phys. Soc.Lond.,(B) "The Plastic Flow of Iron and Plain Carbon Steels above the A_3 Point", Vol.66, p.865, 1953.

5. J.Neil Greenwood and Howard K.Worner,"Types of Creep Curves Obtaines with Lead and Its Dilute Alloys", Journal Inst.Met.,Vol.64#1, 1939, p.135.

6. W.S.Lyman, P.J.Pankaskie and J.E.Dorn, "The Creep Properties of Sand-Cast Aluminum Materials S 5A,GS42A-F and C4A-TG", Alloy Creep 78060, Ser.34,#7,Univ.of Calif. Berkeley, June 1951.

7. A.A.Smith, "Creep and Recrystallization of Lead", Trans.AIME, Vol.143, 1941, p.165.

8. E.S.Machline and A.S.Nowick, "Stress Rupture of Heat-Resisting Alloys as a Rute Process," Trans.AIME, Vol.172, 1947, p.386.

9. W.Kauzman, "Flow of Solid Metals from the Standpoint of the Chemical Rate Theory, "Trans. AIMME, Vol.143, 1941, p.57.

10. S.Dushman, L.W.Dunbar, and H.Huthsteiner, "Creep of Metals." Journal of Applied Physics, Vol.15, 1944, p.108.

11. I.S.Servi and N.J.Grant, "Creep and Stress Rupture Behavior of Aluminum as a Function of Purity," Transactions,AIME, Journal of Metals, Vol.3, p.909.

12. N.J.Grant, and A.G.Bucklin, "On the Extrapolation of Short Time Stress-Rupture Data," Trans.ASM, Vol.42, 1950, p.720.

13. I.S.Servi and N.J.Grant, "Creep and Stress Rupture as Rate Processes", Journal, Institute of Metals, Vol.80, 1951-52, p.33.

14. N.J.Grant, "Creep and Fracture at Elevated Temperatures", Utilization of Heat Resistant Alloys, published by the ASM, 1954.

15. N.J.Grant and A.G.Bucklin, "Creep-Rupture and Recrystallization of Monel from 700 to 1700 F", Trans. ASM, Vol.45, 1953, p.151.

16. M.Herman and N.Brown,"Influence of Order-Disorder Transformation on Creep of Beta Brass", Trans.AIME, Vol.206, 1956, p.604.

APPENDIX I

COMMENTS ON CONSTANT LOAD CREEP CURVES AT 450°C

50,000* Experienced no elongation on loading and slight contraction took place during the first 11 minutes of the run, (the author believes this to be caused by the gage sticking.) Extremely short primary, if any, occurred. Secondary zone lasted for 45 hours.

55,000 Conventional creep curve. Primary ends at 0.1 hours. Insufficient data to accurately determine end of secondary and $\dot{e}_{min}$. From data available $\dot{e}_{min}$ appears to be less than 0.08%/hour.

60,000 Creep curve appears to be conventional. Primary ends at 0.1 hours. Insufficient data to accurately determine end of secondary and $\dot{e}_{min}$. From data available $\dot{e}_{min}$ appears to be less than 0.1%/hour.

65,000 Conventional creep curve.

72,000 Conventional Creep curve.

COMMENTS ON BASIC CREEP CURVE AT 475°C

50,000 Conventional, gage appears to have stuck on loading.

55,000 Conventional, $\dot{e}_{min}$ not accurately established, appears less than 0.08%/hour.

60,000 Conventional, $\dot{e}_{min}$ not accurately determined, appears less than 0.1%/hour.

65,000 Conventional creep curve.

72,000 Conventional creep curve.

COMMENTS ON BASIC CREEP CURVE AT 500°C

25,000 Conventional creep curve. The contraction on loading appears to

* Values refer to initial stress in psi.

have been caused by a stuck gage, as the first five minutes of this run was repeated and a much larger extension was obtained.

30,000 Conventional creep curve. Gage appears to have been stuck for the first two hours, as the first five hours of the run was repeated and a much larger extension was obtained.

35,000 Conventional creep curve.

37,500 Appears to be a conventional creep curve.

39,600 Conventional creep curve for data obtained. Specimen did not break, run ended at 4 hours; the reason for the short run being that the lever arm hit the back support.

40,000 Data for interpretation is scarce. Ferreting an "average" curve through the points, gives a conventional creep curve with $\dot{\delta}_{min}$= 0.42%/hours; time at the start of the secondary stage equals 4 minutes, time of tertiary equals 66 minutes.

42,500 Conventional creep curve.

45,000 Believe gage to be stuck on loading as no elongation was experienced. No primary was exhibited; the tertiary began at 50 minutes.

50,000 Continually accelerating creep curve, although the rate is nearly constant during the first half of the creep life.

55,000 Conventional creep curve.

65,000 Conventional creep curve.

COMMENTS ON BASIC CREEP CURVE AT 525°C

25,000 Creep curve is conventional. Commencement of tertiary cannot be exactly determined because of lack of data. By extrapolation a value of 29 hours appears reasonable.

30,000 No strain versus time data obtained due to slippage of carboloy gage points.

35,000 Indicates a continually accelerating $\dot{e}$

40,000 Exhibits a continually accelerating $\dot{e}$

COMMENTS ON CONSTANT CREEP PLOTS AT 550°C AND 600°C

30,000 Conventional creep curve (550°C)

15,000 Conventional creep curve (600°C). Rupture time not obtained.

20,000 Conventional creep curve. $\dot{e}_{min}$ estimated by extrapolation because of lack of data.

25,000 Conventional creep curve.

30,000 Specimen did not rupture; test ended 0.1 hours, time of rupture extimated 0.15 hours.

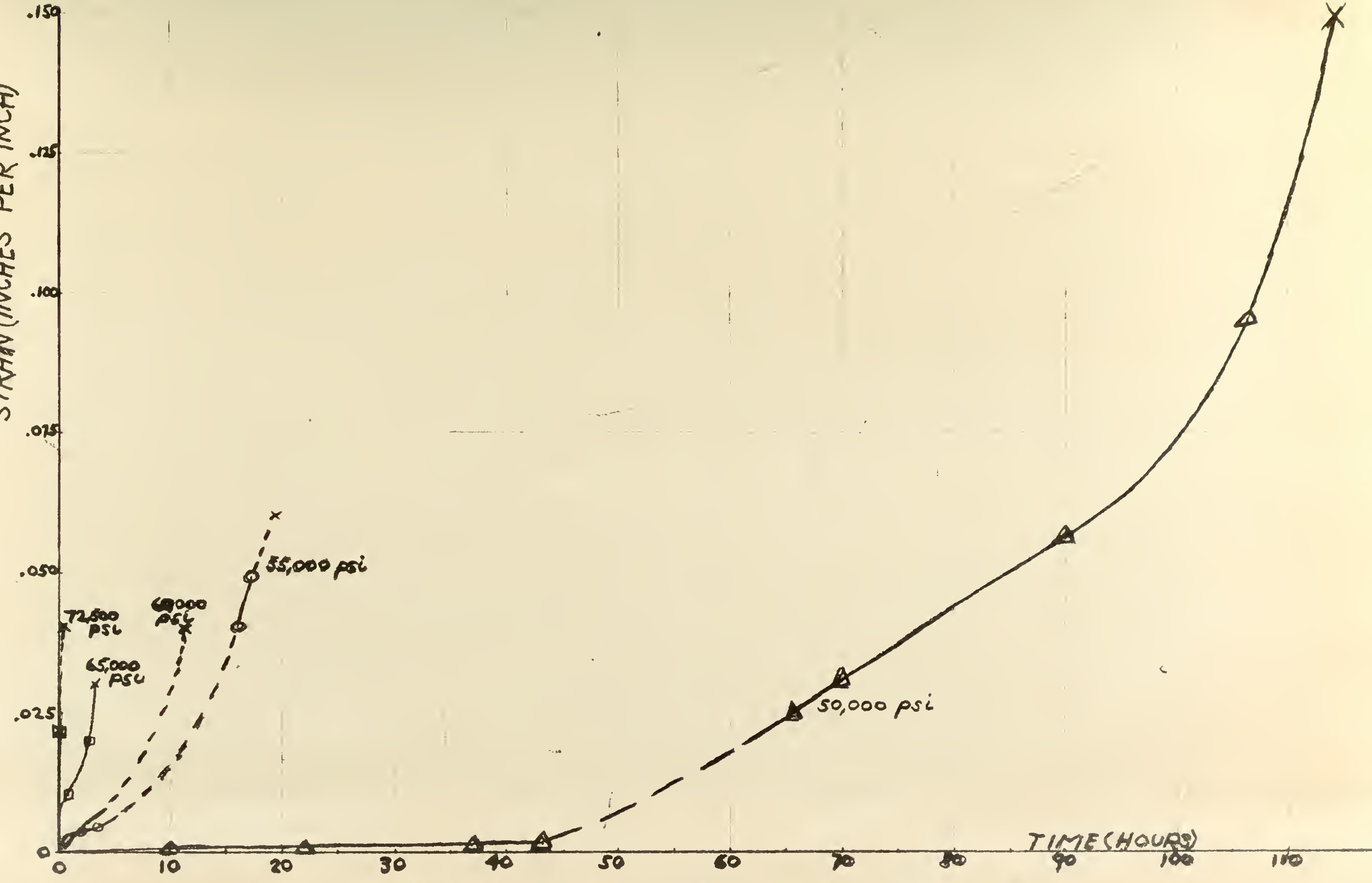


FIG. 8 CONSTANT LOAD CREEP CURVES AT 450°C

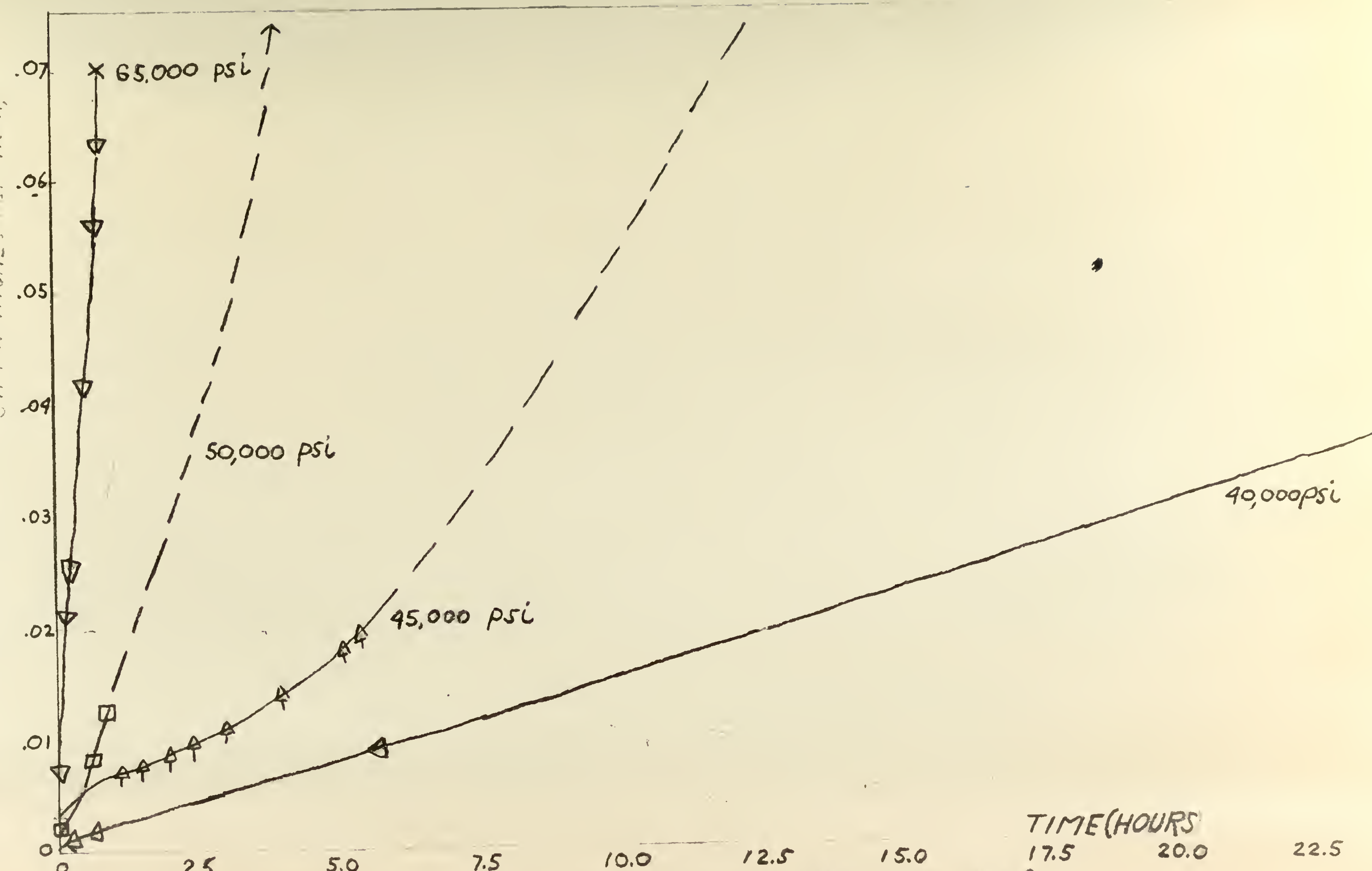


FIG. 9 CONSTANT LOAD CREEP CURVES AT 475°C

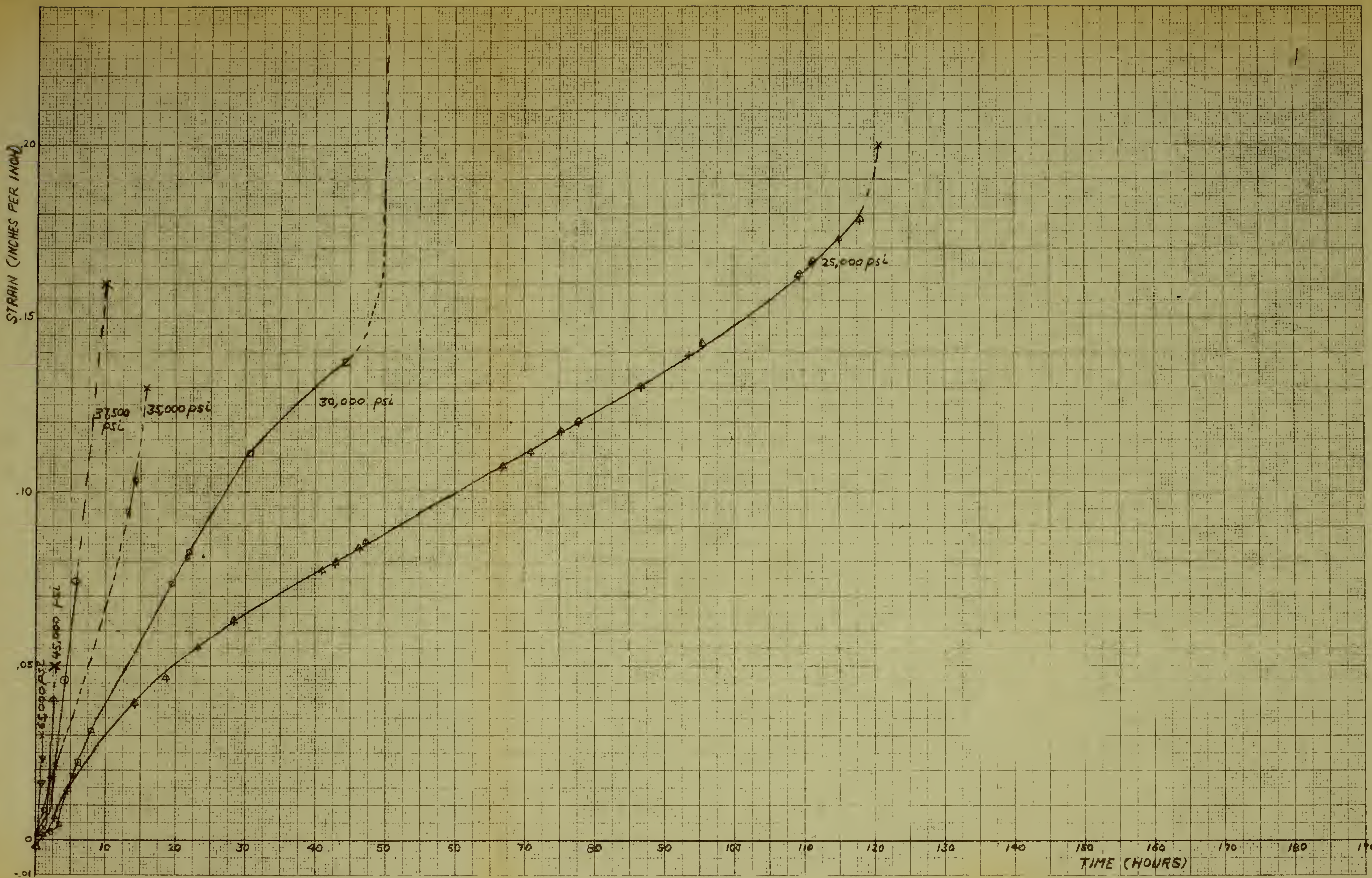


FIG. 10 CONSTANT LOAD CREEP CURVES AT 500°C

FIG. 11 CONSTANT LOAD CREEP CURVES AT 500°C

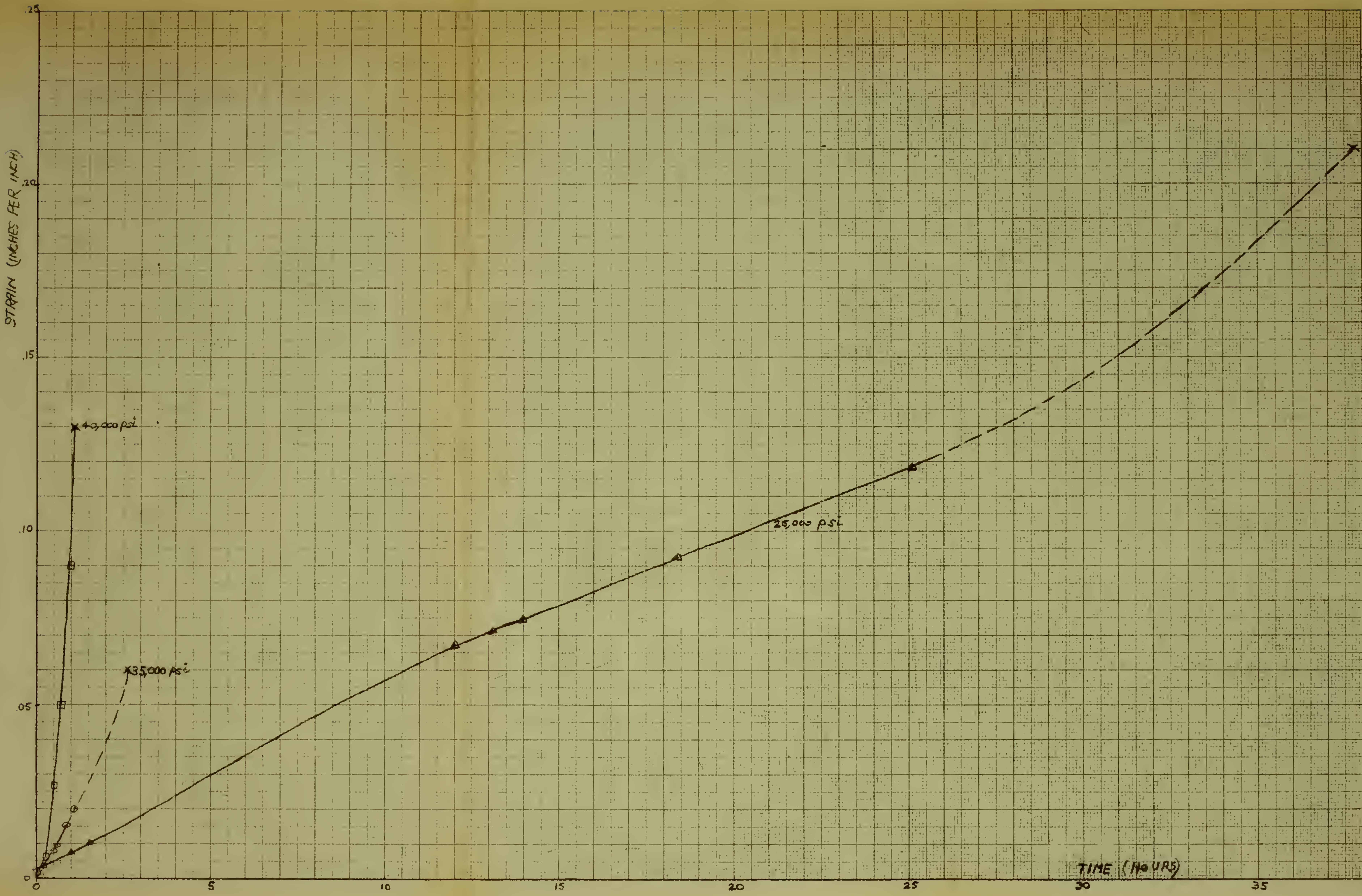


FIG. 12 CONSTANT LOAD CREEP CURVES AT 525°C

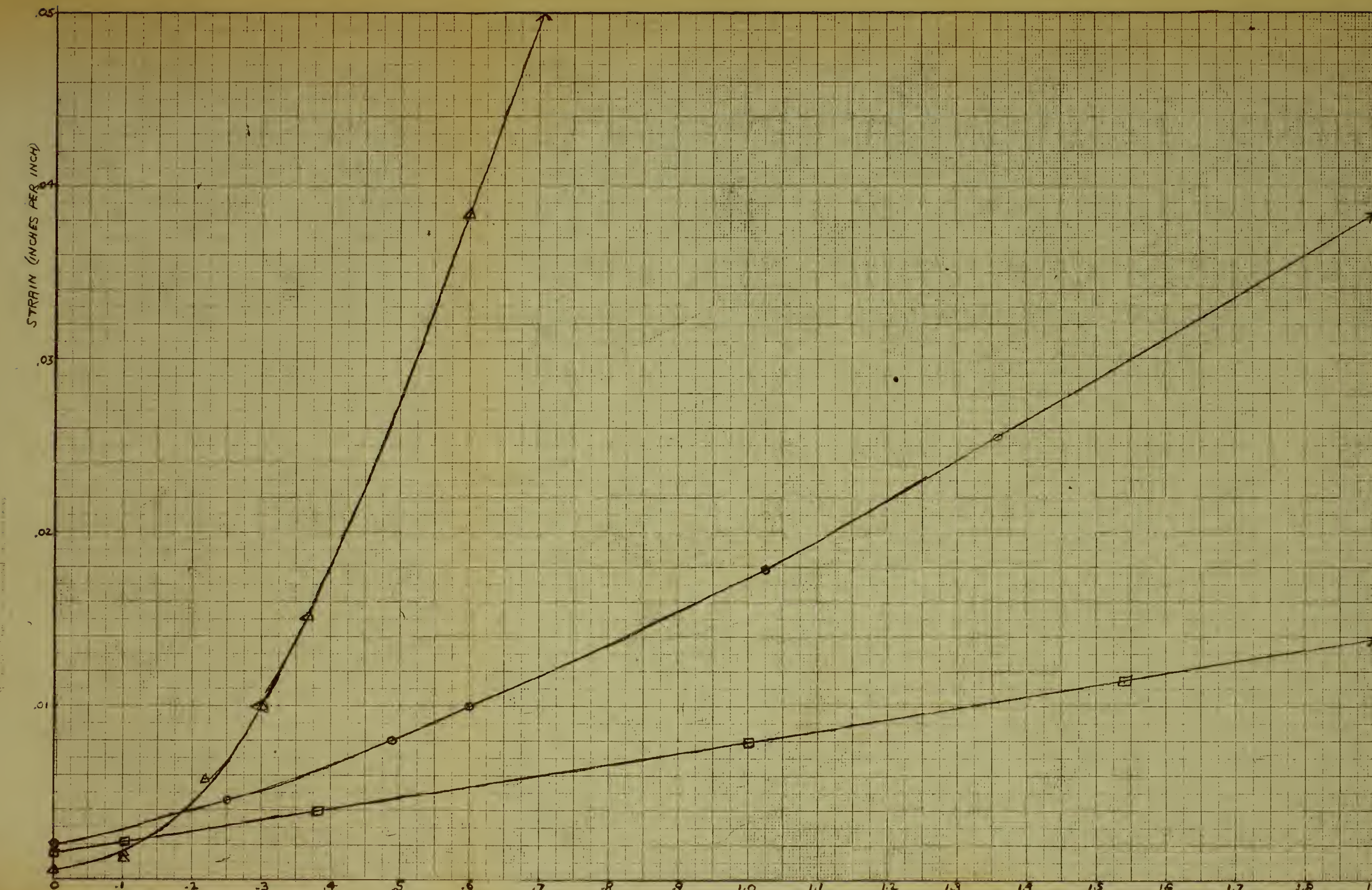


FIG. 13 CONSTANT LOAD CREEP CURVES AT 525°C

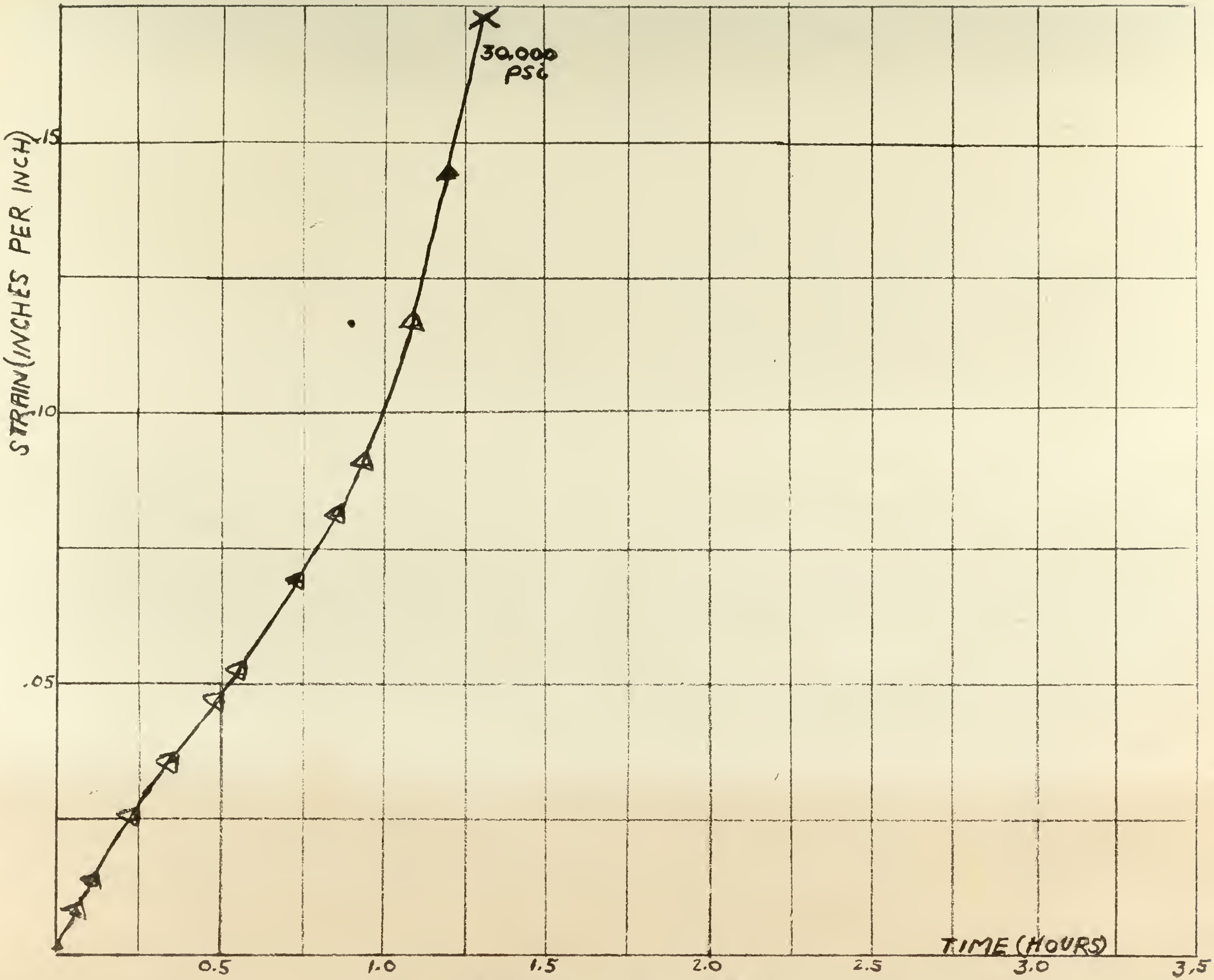


FIG. 14 CONSTANT LOAD CREEP CURVES AT 550°C

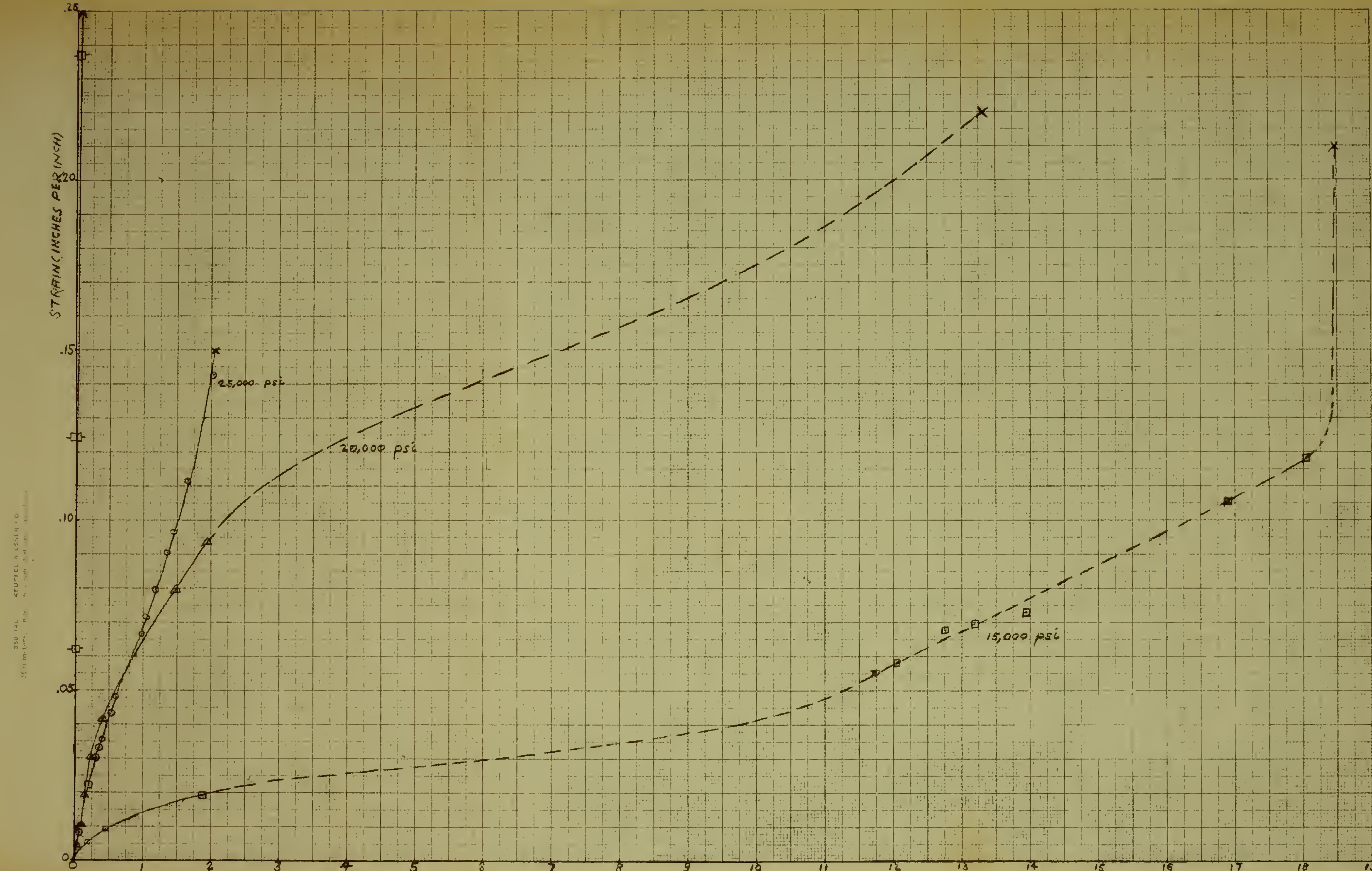


FIG. 15 CONSTANT LOAD CREEP CURVES AT 600°C

APPENDIX II

STRESS	TEMP	TOTAL ELONGATION	TIME TO RUPTURE	MINIMUM CREEP RATE	PRIMARY TIME	TERTIARY START	SECONDARY TIME
PSI	°C	(INS)	(HOURS)	%/HR	HOURS	HOURS	HOURS
72,000	450	.04	.41	6.2	.1	.4	.4
65,000	450	.03	3.29	.25	.1	1.8	.7
60,000	450	.04	11.0	.1	.1	-	-
55,000	450	.06	19.3	.08	.1	-	-
50,000	450	.16	114.1	.001	.1	45	44.9
65,000	475	.07	.79	6.04	.1	.5	.4
50,000	475	.09	4.03	.7100	-	-	-
45,000	475	.08	14.7	.1920	.7	2.9	2.2
40,000	475	.19	180.8	.0465	25	87	62
65,000	500	.03	1.90	.73	.21	.57	.36

APPENDIX II (CON'T)

STRESS PSI	TEMP ^{0}C	TOTAL ELONGATION (INS)	TIME TO RUPTURE (HOURS)	MINIMUM CREEP RATE %/HR	PRIMARY TIME (HOURS)	TERTIARY START (HOURS)	SECONDARY TIME (HOURS)
55,000	500	.03	1.90	.73	.21	.57	.36
50,000	500	.02	2.45	.22	none	0	-
45,000	500	.05	2.70	.12	none	.8	.8
42,500	500	.08		.32	0.2	1.1	.9
40,000	500	.07	3.50	.42	0.05	1.1	
39,600	500	-	-	.2	0.2	1.9	.7
37,500	500	.16	9.91	.35	0.6	1.3	.7
35,000	500	.13	15.42	.54	0.4	2.0	1.6
30,000	500	.28	52.8	.33	31	40	9
25,000	500	.20	120.0	.118	35	87	62.3
48,000	525	.05	.58	-	-	-	-

APPENDIX II (CON'T)

STRESS PSI	TEMP °C	TOTAL ELONGATION (INS)	TIME TO RUPTURE (HOURS)	MINIMUM CREEP RATE %/HR	PRIMARY TIME (HOURS)	TERTIARY START (HOURS)	SECONDARY TIME (HOURS)
40,000	525	.13	1.05	.60	none	none	none
35,000	525	.06	2.62	.996	none	none	-
30,000	525	.22	13.6	-	-	-	-
25,000	525	.21	37.8	.5	12	29	17
30,000	550	.18	1.33	8.04	.31	.59	.28
30,000	600	.33	-	20	-	-	-
25,000	600	.15	2.05	5.25	.3	1.1	.8
20,000	600	.22	13.3	1.00	.4	1.8	1.4
15,000	600	.21	-	.356	3.0	14.2	11.2